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WORK HARDENING IN FACE CENTERED SUBSTITUTIONAL ALLOYS

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WORK HARDENING IN FACE CENTERED SUBSTITUTIONAL ALLOYS

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Abstract

The existing experimental data on the mechanical properties of single crystals of the face centered cubic substitutional alloys is reviewed. It is shown that the data can be understood by supposing that source hardening and impurity hardening are important whereas interaction hardening is not. The free length distribution of dislocations is found to be essentially independent of concentration. A process which is shown to be important for alloys is that by which a new slip band reduces the impurity stresses in its neighborhood. It is suggested that vacancy and interstitial production can explain the results. Several experiments are suggested.

Experiment

Stress strain data exists for single crystals of a brass⁽¹⁾, silver gold⁽²⁾, and copper nickel alloys⁽³⁾ over a range of concentrations. The curves obtained by Sachs and Meerts⁽²⁾ for the silver gold system are shown in Figure 1. The stress strain curves appropriate for the other alloys mentioned are similar.

There are several other pieces of experimental evidence which are available: Goler and Sachs⁽⁴⁾ and later Treuting and Brick⁽⁵⁾

* Supported by the O.N.R.

have shown that slip lines cluster in a brass, i.e. that additional deformation is accomplished by glide on slip bands which are near previously formed slip bands. Figure 2 shows this behavior according to Treuting and Brick.

There is also data on asterism⁽⁶⁾ which reveals that in alloys, in contrast to pure metals, very little asterism occurs until slip on a second slip system begins.

The change in the orientation of a tensile specimen during glide is such that it reduces the resolved shearing stress on the initial slip system so that eventually the resolved shearing stresses on two slip systems become equal. In pure metals further deformation occurs by simultaneous equal amounts of glide on the two slip systems⁽⁷⁾. The alloys on the other hand glide on only one slip system at a time⁽⁷⁾. There is also in alloys the phenomena of "overshooting" i.e. the alloys continue to deform on the initial slip system past the orientation where the two resolved shearing stresses are equal⁽⁶⁾. In fact a careful examination of the data indicates that the deformation continues on the initial slip system until the resolved shearing stress on the second slip system exceeds that on the first by the yield stress of the alloy. This conclusion is illustrated in Figure 3 by data on a brass showing the discontinuous increases in the resolved shearing stress which accompany the transition of slip from the initial slip system to the second system.

Theory

The fact that little or no asterism occurs until deformation on the second slip system has taken place leads us to the conclusion that source hardening⁽⁸⁾ and not interaction hardening is most important for these alloys. In the case of source hardening the longest free lengths of dislocation generate and become locked. An increase in stress is then necessary to produce Frank-Read⁽⁹⁾ generation at the shorter dislocations. Interaction hardening was first discussed by G. I. Taylor⁽¹⁰⁾. The basic idea is that the larger the dislocation density the greater will be the applied stress necessary to drive dislocations through the existing tangled skein of dislocations.

The Treuting and Brick pictures show that source hardening occurs in a brass at small strains. This is evident because only finite glide occurs at each slip band. Their pictures also show that some process occurs near a slip band which makes it easier for a second slip band to be produced in the vicinity of the first. It will be proposed here that this process reduces the impurity hardening in the vicinity of the original slip band. It will be seen later that this hypothesis explains the phenomena which take place when a change of slip system occurs.

The applied stress σ necessary for deformation is therefore:

$$\sigma = \sigma_1 + \sigma_s$$

where σ_1 is the portion of the applied stress necessary to overcome the influence of the impurities and σ_s is the portion of the applied stress needed for Frank-Read generation.

At this point a second assumption will be made. In Figure 1 the slopes of the ascending straight line portions of the stress strain curves agree to within 20 % of their value. It will therefore be assumed that the dislocation free length distribution is independent of impurity concentration. The data on a brass indicates that there the source hardening may be slightly decreased by increasing the impurity concentration.

There is still a point which has not been taken care of. Experimentally the data shows that the linear portion of the stress strain curve does not extrapolate back to zero strain when the stress is zero. Two possible explanations can be offered. It is possible that the presence of impurities alters the locking of sources in such a way that the average glide per slip band is larger in the flat portion of the stress strain curve than it is in the ascending portion where the impurity stresses cease to be important for glide on the initial slip system. If this is the case it can be checked by direct observation of the glide at the slip bands in the first and second stages of deformation. Another possibility is that the cloud of dislocations which move through the lattice at yielding sweep some of the shorter dislocations along with them. This alternative would predict that the transition from the flat to the ascending curve would not be sharp, but that the initial portion of the ascending curve would require larger stresses than otherwise anticipated because some of the short dislocation lengths have already been used. The data shown in Figure 1 does show a transition region of the sort predicted by the second alternative.

It should be mentioned however that inhomogeneities in concentration could also cause such a transition region.

In the theoretical paper⁽⁸⁾ on source hardening in pure metals an expression for the free length distribution was obtained by fitting experimental data valid for aluminum. Such a fit is not illuminating in the case of the alloys since there are more constants to be determined than there are pieces of experimental information. However by just blindly using the dislocation distribution found for aluminum the slope calculated for silver gold is too small by a factor of three. Thus the distribution must be nearly the same as that found for pure aluminum.

There is still the transition from one glide system to another to be considered. According to the picture presented here glide on the first slip system eventually completely erases the impurity stresses for further glide on that slip system. The experimental findings on overshooting mentioned earlier show that glide on the second slip system must still overcome the impurity stress as well as the source stress. This result indicates that the nullification of impurity stresses occurs only in the vicinity of existing slip bands. Thus in Figure 4 the effect can aid slip on the initial slip system, but since glide on a new slip system must cut through material far from any existing slip band no aid can be expected.

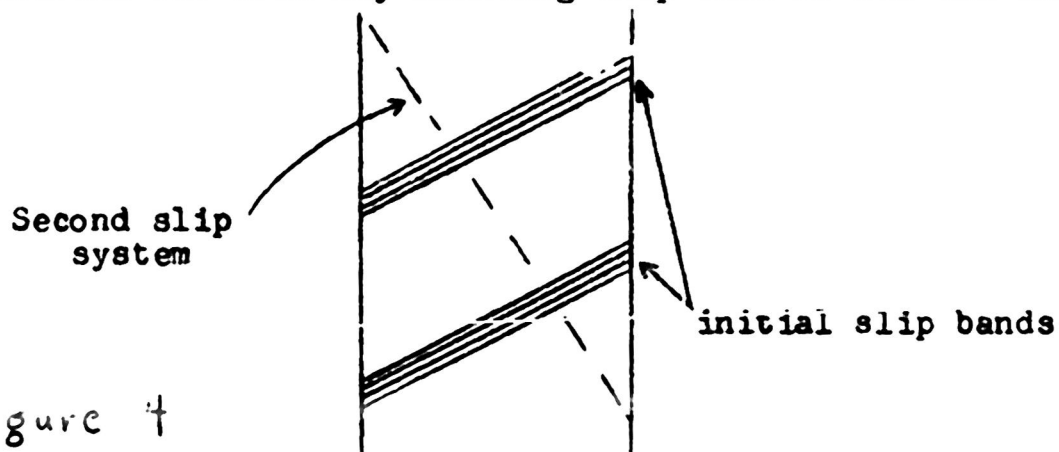


Figure 4

It would be of interest to know in this connection whether the distance over which nullification is effective depends on temperature. Seitz has suggested⁽¹¹⁾ that the glide on a slip band produces vacancies and interstitials there. These could move about a bit and then settle down in such a way as to erase the local impurity stresses. They prefer to erase local stresses because in so doing they lower the total potential energy of the system. At low temperature they may still be able to move a bit because of the excess local vibrational energy in the vicinity of the glide band. Even so it would be valuable to deform a single crystal of some alloy at low temperature to make certain that a reduction of the range of this "purification" effect does not take place. If enough reduction in range occurs so that new slip lines cannot form near the old ones because of the large local dislocation stresses (i.e. if the range drops below a few hundred Angstroms) two effects will occur. First, there will be no flat portion of the stress strain curve. Second, there will be no "overshooting". Slip band separation should also be measured after a small amount of deformation at the low temperature to determine whether a reduction in the range has resulted in more closely spaced slip bands. It is probable that deformation studies at high temperatures would be more useful. Increasing the temperature of deformation may lead to more widely spaced slip bands. In addition the flat portion and the "overshooting" might show changes.

There is a further prediction which can be made. If the theory given in the source hardening paper for the temperature

dependence of the initial stress strain curves of aluminum is correct. Then it follows that any alloy which possesses complex slip bands containing several glide lamellae will also have stress strain curves in which the ascending straight line portions depend on temperature. Certain aluminum alloys may well have stress strain curves which depend on temperature in the fashion just described.

There are two points concerning the change in slip systems which should be mentioned. First, if the specimen is so oriented that double slip would occur in a pure metal at a strain which is within the flat portion of the stress strain curve of the alloy then since "purification" on the initial slip system is incomplete one would not expect the differences in the resolved shearing stresses to be as much as the yield stress when slip changes from one system to another. Second, it can be seen as follows why slip continues to use only the second slip system once it has begun there: To begin with the dislocation sources on all slip systems are gradually used as glide proceeds.⁽⁸⁾ It is only those on the initial slip system which produce an appreciable amount of glide and which generate large numbers of vacancies and interstitials. In the initial slip system the distribution is used until at the strain associated with the change in slip system the longest length l_1 available is given by:

$$\sigma_1 = \frac{Gb}{l_1}$$

where G is the modulus of rigidity, b is the Burgers vector of the

dislocation, and σ_1 is the resolved shearing stress in the initial system. Similarly for the second system:

$$\sigma_2 = \sigma_1 + \sigma_y = \frac{Gb}{l_2} + \sigma_y$$

Thus $l_2 = l_1$ so that the dislocation distributions are used to the same extent when the change occurs. But when slip occurs on the second system the purification process begins there. In addition, some of the vacancies and interstitials generated by the glide on the second slip system will be deposited in the region where additional slip would occur on the first slip system. This acts just as an impurity would and hardens the initial slip system where complete local nullification has previously occurred. The vacancy interstitial processes thus simultaneously purify the second slip system and add impurities to the initial slip system. This implies that when the orientation changes until slip again takes place on the initial slip system the resolved shearing stress on the initial slip system will be larger than that on the second, but that the difference will be determined by vacancy interstitial stresses not by impurity stresses.

The principal conclusions of this discussion are: First, that some process occurs during glide in face centered cubic alloys which is able to erase the impurity stresses in the vicinity of the active slip bands. It is suggested that this process may be the production, the motion, and the settling down of vacancies and interstitials in such a way as to nullify impurity stresses. Second, it is suggested that source hardening and impurity hardening

play major roles in these alloys whereas interaction hardening is rather unimportant for their mechanical properties. Third, it is found that the free length dislocation distribution is independent of concentration and in fact does not seem to vary much from one metal to another. Fourth, it is felt that much basic information can be gained from further study of these alloys and to this end a number of experiments are suggested.

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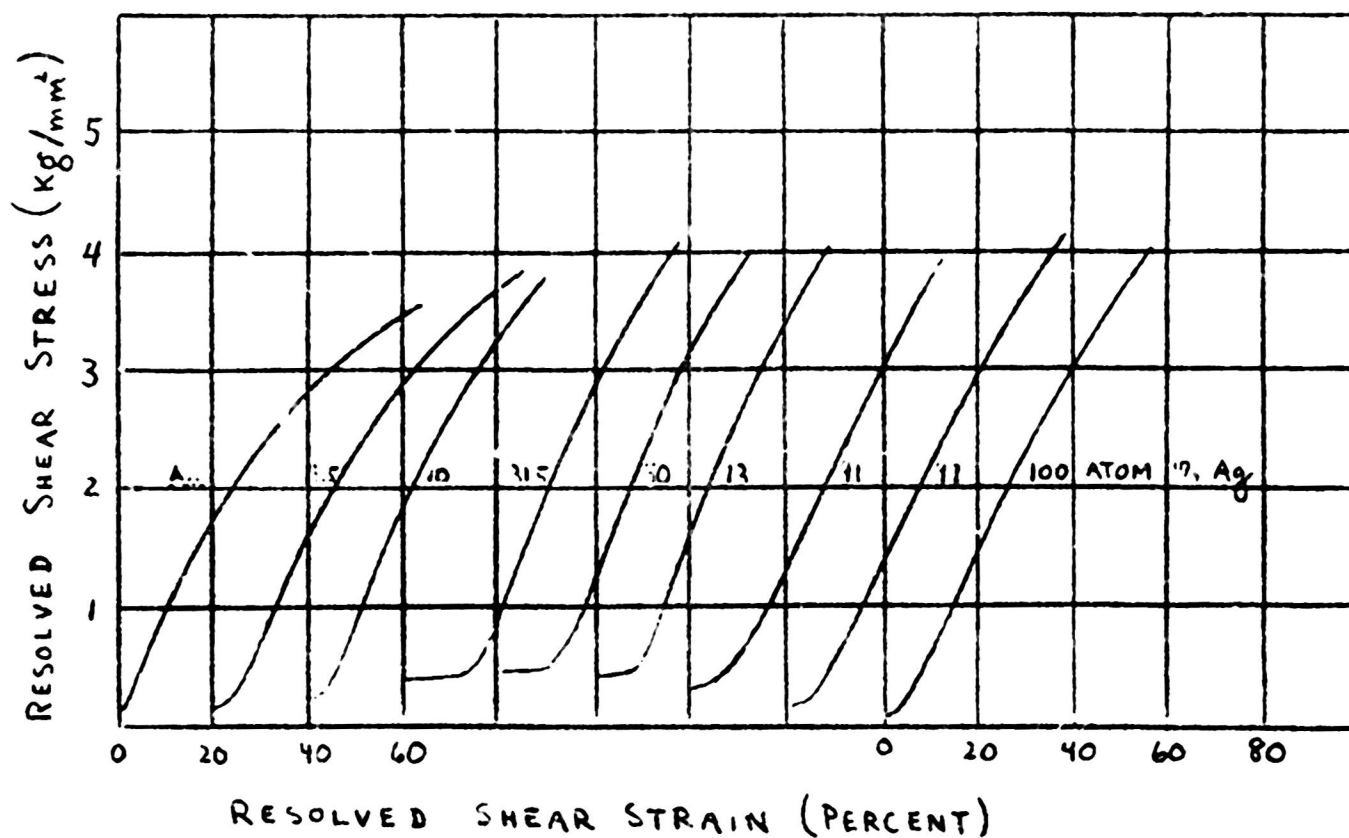


Figure 1. Resolved Stress Strain Curves for Silver Gold single Crystals of Various Compositions (after Sachs and Weerts).

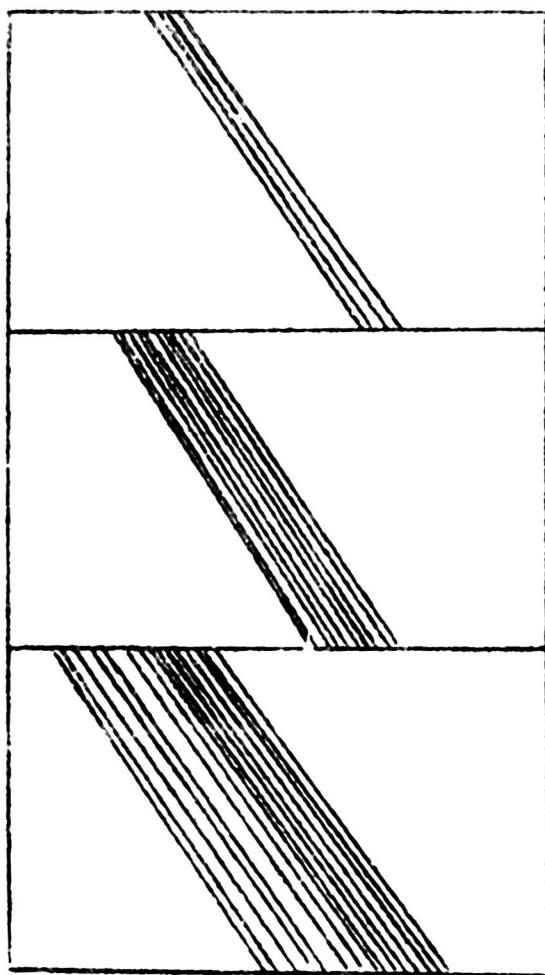


Figure 2. Primary Slip Steps in 72 % Single crystals of α Brass. Upper Picture 0.002 shear, Middle Picture 0.004 shear, Lower Picture 0.015 shear. (After Treuting and Brick)

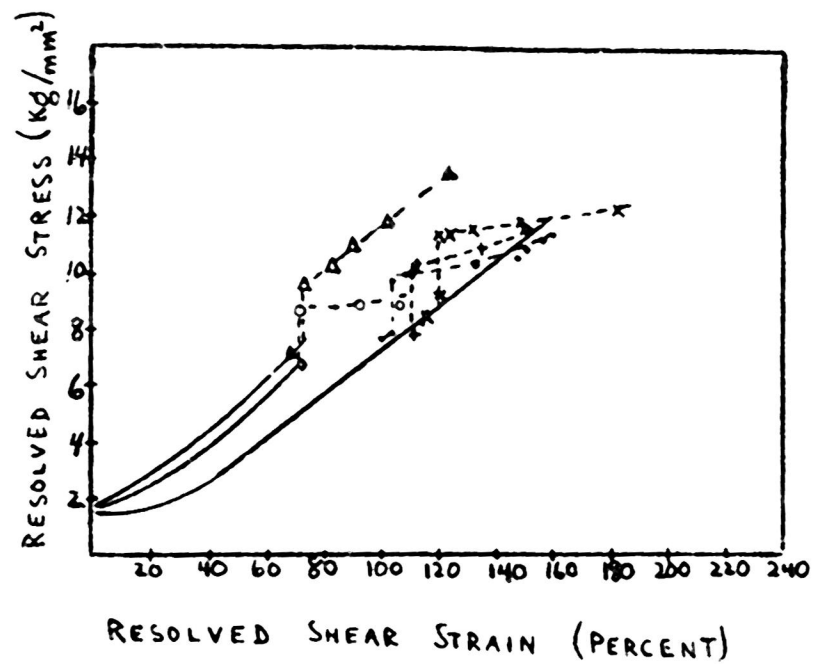


Figure 3. Resolved Stress Strain Curves for Various Single Crystals of 72 % α Brass. Breaks Occur at the Onset of Slip on a Second Slip System. (After Masima and Sachs).